Elevated Temperature Deformation Behavior of Nanostructured Al-Ni-Gd Alloys

X. L. Shi[1], R. S. Mishra[1] and T. J. Watson[2]

[1]Department of Metallurgical Engineering, University of Missouri, Rolla, MO 65409
[2]Materials and Processes Engineering, Pratt & Whitney, East Hartford, CT 06108

Keywords: Aluminum alloy; deformation behavior; dispersion strengthening; nanostructure; threshold stress

Abstract

Crystallization of amorphous alloys provides opportunities to produce bulk nanostructured materials with high volume fraction of second phase particles. In this study, the elevated temperature deformation behavior of $Al_{87}Ni_7Gd_6$ alloys in which rod-like nano-crystalline particles are dispersed along the grain boundaries was characterized in the temperature range of 523 to 673 K. The results, which cover four orders of magnitude in strain rate, show an increase in apparent stress exponent with decreasing temperature. The introduction of a threshold stress into the analysis leads to stress exponent of ~5 and a true activation energy of 136 kJ/mol in $Al_{87}Ni_7Gd_6$ alloy. The normalized threshold stress exhibits strong temperature dependence. The operative deformation mechanism is discussed in terms of dislocation climb.

1. Introduction

Development of high specific strength aluminum alloys that are microstructurally and mechanically stable at temperature up to 573 K has been pursued for the last two decades for elevated temperature applications [1]. For high temperature applications, the classical precipitate strengthening has the disadvantage of losing it strengthening effect due to the coarsening/dissolution of precipitates. So thermally stable dispersoids, such as oxides, carbides and intermetallic compounds, are useful to maintain the strengthening at high-temperatures. Recently, development of the high-strength Al-based alloys via non-equilibrium processing has received significant attention. In particular, crystallization of amorphous alloys provides opportunities to produce bulk nanostructured materials with high volume fraction of second phase particles. The phase transformation from amorphous state has some unique features: (1) homogenous nucleation, (2) high nucleation frequency, (3) low growth rate, and (4) nanoscale inter-particle spacing [2]. All of these are beneficial for developing high-strength dispersion-strengthened alloys.

In this paper, the elevated temperature deformation behavior of as-extruded dispersion-strengthened $Al_{87}Ni_7Gd_6$ (at %) alloy was evaluated at various temperatures and strain rates. In addition, the observed deformation behavior is compared to results of a number of dispersion- strengthened aluminum alloys with metallic/intermetallic dispersoids.

2. Experimental procedure

The atomized $Al_{87}Ni_7Gd_6$ powders were consolidated at 422 K, and extruded into 22.9 mm (0.9 inch) diameter rods at 623 K with an extrusion ratio of 11:1. Mini-tensile specimens (gage length 1.3mm and gage width 1.0mm) were electro-discharge machined from the as-extruded rods in the longitudinal direction. The tensile specimens were polished to the final thickness of ~0.5 mm and 1 μm finish. Tension tests were performed on a custom-built, computer-controlled mini-tensile tester in the temperature range of 523 to 673 K and initial strain rates range of $5x10^{-5}$ to $1x10^{-2}$ s^{-1}.

The microstructure of the extruded rod was examined using a Philips EM430 transmission electron microscope (TEM) at 300 kV. TEM specimens were prepared using ion-milling.

3. Results

Figure 1 shows a TEM micrograph of as-extruded $Al_{87}Ni_7Gd_6$ alloy. Rod-like particles with an average length of ~180 nm and diameter of ~30 nm are dispersed along the grain boundaries. The average grain size is ~200 nm.

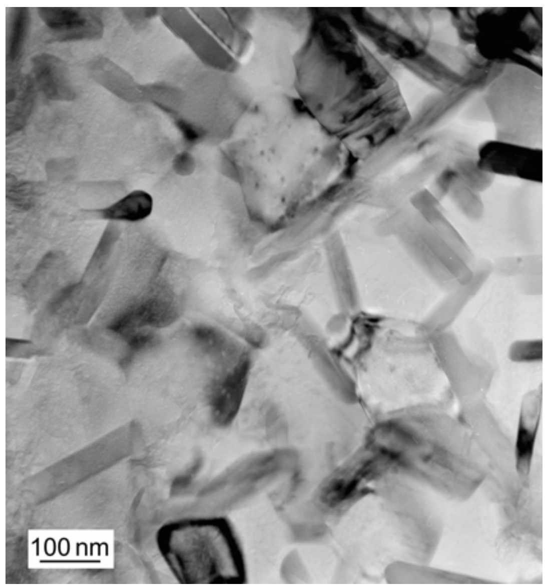

Fig 1. Bright-field TEM micrograph of as-extruded $Al_{87}Ni_7Gd_6$ alloy.

The variation of tensile strength with the temperature at strain rate$1x10^{-3}s^{-1}$ is shown in Fig. 2. It can be noted that the tensile strength of this alloy was 445MPa at 523 K. True stress-strain curves at various strain rates at 673 K are shown in Fig. 3. At 673 K, the as-extruded alloy exhibited good ductility at all strain rates. However, there is no apparent relationship between strain rate and failure strain. The maximum failure strain of ~ 0.32 was obtained at strain rate of $1x10^{-4}$ s^{-1}. The variation of flow stress with strain rate is shown in Fig. 4. The apparent stress exponent increases with decreeing temperature.

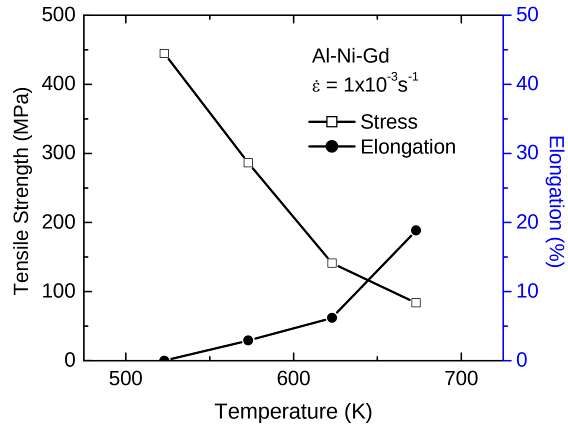

Fig.2.Variation of tensile strength with temperature at initial strain rate of $1\text{x}10^{-3}\text{s}^{-1}$for $Al_{87}Ni_7Gd_6$ alloy.

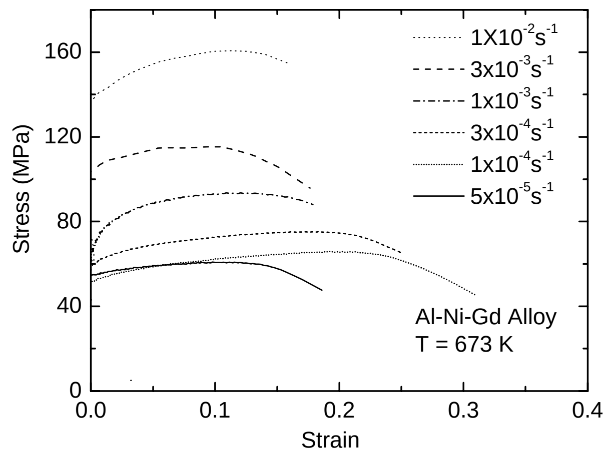

Fig.3. Stress-strain behavior of $Al_{87}Ni_7Gd_6$ alloy at 673 k for various initial strain rates.

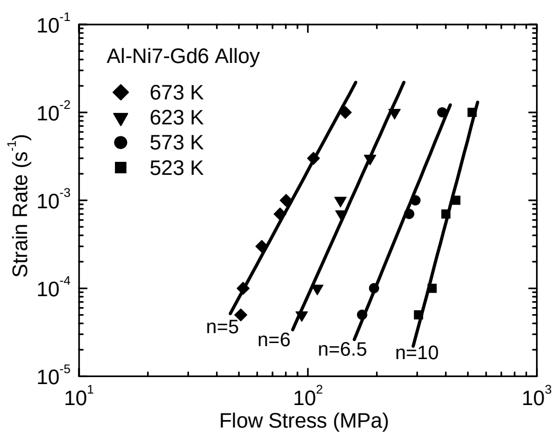

Fig.4. Double logarithmic plot of strain rate versus flow stress at different temperatures.

4. Discussion

4.1 Microstructure

It is well recognized that significant strengthening of Al-based can be achieved by utilizing non-equilibrium processing [3-4]. For high temperature applications, the focus is to produce alloys with dispersion strengthening and nanoscale microstructures. As shown in Fig.1, the intermetallic compounds do not exhibit preferred orientation. It is apparent that hot extrusion of the amorphous powders can result in the formation of ultra-fine structure consisting of fine ternary compound ($Al_{15}Ni_3Gd_2$) and binary compound M_3Gd (M-Al, Ni) [5], embedded in the Al matrix.

4.2 Elevated temperature deformation behavior

The tensile strength of this alloy was 445 MPa at 523 K at strain rate of $1x10^{-3}$ s^{-1}. The good elevated temperature strength is due to the dispersion of the intermetallic compounds which are thermally stable and insoluble in the matrix and act as the barrier to the dislocation movement.

The observed stress exponents are higher than 5. In order to rationalize the high stress exponents in dispersion-strengthened alloys, the threshold stress concept is often introduced [6]. Lagnegorg and Bergman [7] introduced the widely used method to determine threshold stress by plotting $\dot{\varepsilon}^{1/n}$ versus σ and extrapolating the linear fitted line to zero $\dot{\varepsilon}$. The present data fits well with the true stress exponent of 5, shown in Fig. 5. It can be seen that threshold stress varies with temperature. Similar trend of temperature-dependent threshold stress has been observed in other dispersion-strengthened Al-based alloys, as is shown in Fig. 6 [8-11]. In the present work, the apparent activation energy in the temperature ranges of 573 to 673 K was estimated to be 207 kJ/mol. The high apparent activation energy is rationalized to 136 KJ/mol after the temperature dependence of the threshold stress is considered, as is shown in Fig. 7. This value agrees well with activation energy for volume diffusivity in pure Al (Q_L = 142 KJ/mol). The threshold stress model proposed by Mishra et al [12] suggests that the origin of threshold stress is due to the attractive dislocation-particle interaction. This model as well as other models for threshold stress for dislocation creep is based on interaction of lattice dislocations with particles within the grain. However, as noted earlier, the present alloy has particles mostly on the grain boundaries. For such nanostructured materials, the theoretical framework for dislocation-particle interaction has not been developed.

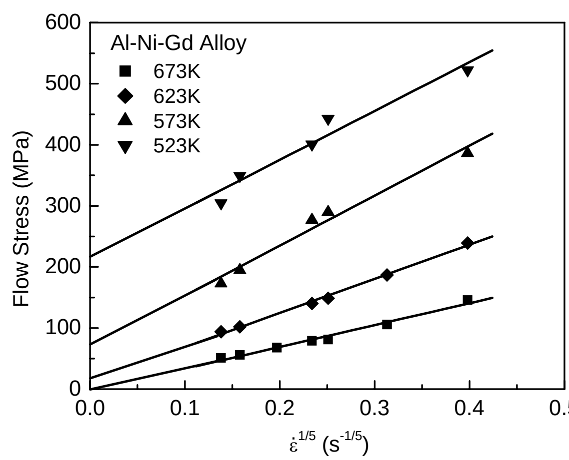

Fig.5. The variation of flow stress as a function of $\dot{\varepsilon}^{1/5}$ in $Al_{87}Ni_7Gd_6$ alloy.

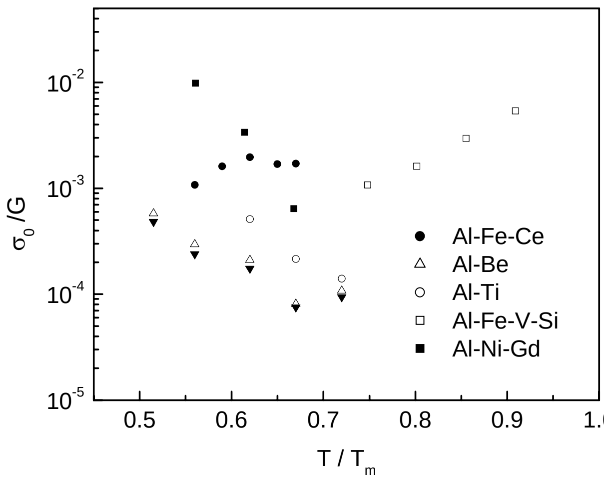

Fig.6. Threshold stress (normalized with respect to the shear modulus) as a function of homologous temperature for dispersion-strengthened alloys

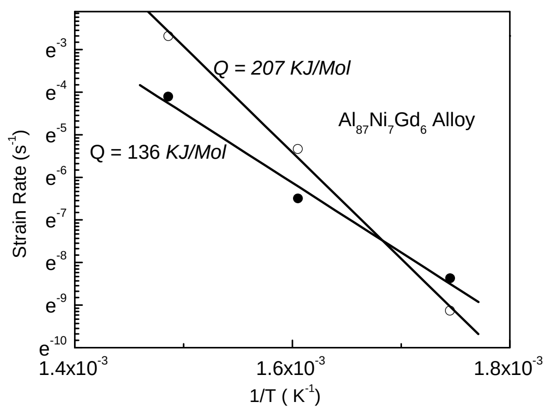

Fig. 7. An Arrhenius plot of strain rate against inverse of temperature

The deformation behavior of a material at elevated temperatures can be represented by the constitutive equation which incorporates threshold stress as given by [13, 14]

$$\dot{\varepsilon} = (AGbD_L)/(kT)\ ((\sigma-\sigma_0)/G)^{\,n}$$

where $\dot{\varepsilon}$ is the strain rate, D_L the lattice self-diffusivity, G the shear modulus, b the Burgers vector, k the Boltzmann's constant, T the absolute temperature, σ the applied stress, σ_0 the threshold stress, A the dimensionless constant, n the true stress exponent. The normalized strain rate, $\dot{\varepsilon}kT/DGb$, is plotted against the normalized effective stress $(\sigma-\sigma_0)/G$, on double logarithmic scales in Fig. 8. For the data analysis, values of the lattice self-diffusivity were obtained using D_L [M^2/s] = 1.71 x 10^{-4} exp (-142/RT) [15], and G [MPa] = 3.0 x 10^4-16T [16], where R is the universal gas constant. It can be seen that the slope of the fitted line is 5, which indicates the dislocation-climb controlled deformation. For comparison, a number of dispersion-strengthened alloys with metallic/ intermetallic dispersoids and pure Al are also included in this plot [9-12,17]. The nanostructured alloy in the present work exhibits good elevated temperature strength, which is superior to that for conventional Al-based alloys as well as for Al-based alloys developed by rapid solidification processing. The achievement of good elevated temperature strength of this alloy is presumably because of the formation of the finely mixed nanostructure consisted of high volume fraction of thermal stable second phase particles homogenously distributed in the Al matrix, which cannot be produced by the conventional thermo-mechanical processing. In this case, the high thermal stability of the nanoscale second phase particles is presumably due to the absence of the internal defects in the particles due to the unique fabrication route [2] and slow diffusivity of Gd and Ni.

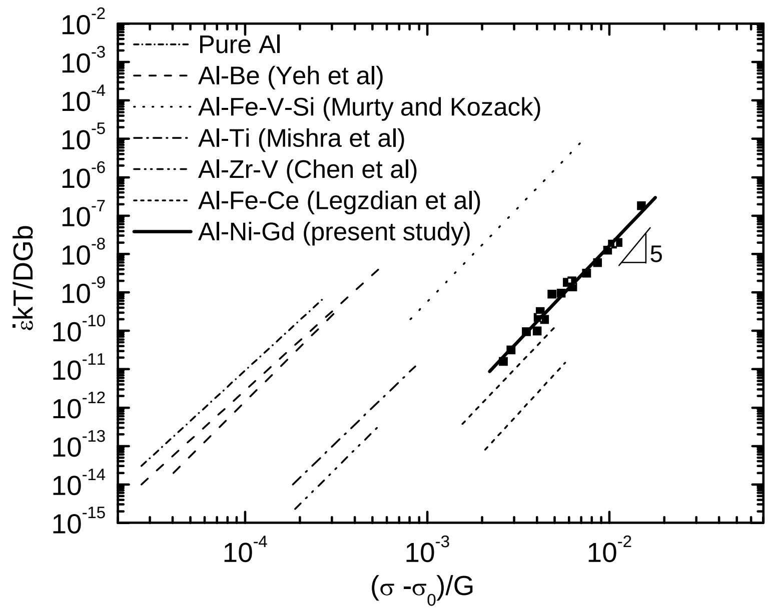

Fig .8 Temperature and diffusivity compensated strain rate versus normalized effective stress for dispersion-strengthened materials.

5. Conclusions

- Homogenous dispersion of fine intermetallic compounds can be obtained by the use of the amorphous precursor. Nanostructured $Al_{87}Ni_7Gd_6$ alloy exhibits good elevated temperature strength.
- The apparent stress exponent was higher than 5 and the apparent activation energy was 207 KJ/mol. The $Al_{87}Ni_7Gd_6$ alloy exhibits temperature-dependent threshold stress.
- The Deformation mechanism is likely to be dislocation-climb mechanism.

6. Acknowledgment

The authors gratefully acknowledge the financial support of DARPA through contract number F336115-01-2-5217.

References

1. Y-W. Kim and W.M.Griffth (eds), *Dispersion strengthened aluminum alloys* (TMS-AIME, Warrendale, PA, 1988), 31

2. A. Inoue, H.M. Kimura, "Fabrication and Mechanical Properties of Bulk Amorphous, Nanocrystalline, Nanoquasicrystalline Alloys in Aluminum-based System" *J. Light Metals*, 1 (2001), 31-41
3. J.S. Benjamin and M.J. Bomford, "Dispersion Strengthened Al Made by Mechanical Alloying" *Metal. Trans. A*, 8A (1977), 1301-1305.
4. F.H. Froes, C.M. Ward-Close and W. Baeslack, "Advanced Synthesis of Light-Weight Metals" *J. Adv. Mats*., 25(1)(1993), 20.
5. M.C. Gao and G.J. Shiflet, "Devitrification Phase Transformations in Amorphous Al sub 85 Ni sub 7 Gd sub 8 Alloy" *Intermetallics*, 10(11-12)(2002), 1131
6. J.H. Gittus, "Theoretical Equation for Steady-state Dislocation Creep in a Material Having a Threshold Stress," *Proc. Roy. Soc. A*, 342(1975), 279-287.
7. R. Lagneborg, and B. Bergman, "Stress/Creep-Rate Behavior of Precipitation-Hardened Alloys" *Metal. Sci*.. 10(1976), 20.
8. D. Legzdina, T.A. Parthasarathy, "Deformation Mechanisms of a Rapidly Solidified Al--8.8Fe--3.7Ce Alloy" *Metall. Trans. A*, 18A (10)(1987), 1713
9. Y.H. Yeh, N. Nakashima, H. Kurishita, S. Goto, H. Yoshinaga, "Threshold Stress for High-Temperature Creep in Particle Strengthened Al--1.5 vol.% Be Alloys" *Mater. Trans, JIM*, 31(4)(1990), 284
10. R.S. Mishra, A.G.Paradkar, K. N. Rao, "Steady State Creep Behavior of a Rapidly Solidified and Further Processed Al--5 wt.% Ti Alloy" *Acta Metall.Mater*.41 (1993), 2243
11. G.S. Murty, M.J. Koczak, "Rate Sensitivity in the High Temperature Deformation of Dispersion Strengthened Al--Fe--V--Si Alloys" *J. Mater. Sci*., 24(2)(1989), 510
12. R.S. Mishra, and T.K. Nandy, "The Threshold Stress for Creep Controlled by Dislocation--Particle Interaction" *Phil. Mag. A*, 69 (1994), 1097-1109
13. K.T. Park, E. J. Lavernia, F.A. Mohamed, "High Temperature Creep of Silicon Carbide Particulate Reinforced Aluminum" *Acta Metall.Mater*.38 (1990), 2149
14. K.T. Park, E.J. Lavernia, F.A. Mohamed, "High-temperature Deformation of 6061 Aluminum" *Acta Metall.Mater*.42 (1994), 667
15. J.E. Bird, A.K. Mukherjee, J.E. Dorn, in: D.G. Brandon, A. Rosen (Eds.), *Quantitative Relation between Properties and Microstructure* (Israel University Press, Jerusalem, 1969), 225
16. T. S. Lundy, J. F. Murdock, *J. Appl. Phys*. 33(1962), 1671
17. Y.C. Chen, M.E. Fine, J.R. Weertman, "Microstructural Evolution and Mechanical Properties of Rapidly Solidified Al--Zr--V Alloys at High Temperatures" *Acta Metall.Mater*.38 (1990), 771